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Review

Tribological behaviour of C/C–SiC composites—A review

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Abstract: Ceramic matrix composites made of carbon fibres and carbon matrix (C/C) are generally used for aircraft structures and brake discs due to their low density, and good thermal, mechanical, and tribological properties. Silicon carbide (SiC) can be introduced to the matrix to improve the performance of C/C composites, because it increases the hardness and thermal stability, and decreases the chemical reactivity, which leads to the improvement of tribological properties of C/C composites. Thus carbon–carbon silicon carbide (C/C–SiC) composites can be used at high temperature for the application of brake discs, friction clutches, etc. C/C–SiC composites are fabricated by three different methods: (i) chemical vapour infiltration (CVI), (ii) polymer infiltration and pyrolysis (PIP), and (iii) liquid silicon infiltration (LSI), among which LSI method is widely used for the fabrication of C/C–SiC composites due to higher mechanical and thermal properties.

Keywords: carbon–carbon (C/C) composites; coefficient of friction; wear

1 Introduction

Carbon fibre reinforced carbon (C/C) composites have unique properties like high melting point, high stiffness and toughness, better thermal shock properties, low coefficient of thermal expansion, etc., due to which C/C composites are used in aircraft brake applications, racing cars, and motorcycles [1–5]. C/C composites have high wear rate and poor oxidation resistance which limit its application [6,7].

The introduction of silicon carbide (SiC) to the matrix of C/C composites leads to the application at high temperature and the improvement in tribological properties due to the chemical stability and hardness of SiC and self-lubricating action of carbon. The mechanical properties of carbon–carbon silicon carbide (C/C–SiC) composites are not affected at high temperature [8,9]. C/C–SiC composites have lower density (about 2.0 g/cm^3), longer service life, and higher thermal shock resistance as compared to traditional brake materials. When the weight percentage of silicon carbide is higher than 20% in the C/C–SiC composite, the composite shows lower sensibility to surroundings and temperature [10].

Research is going on C/C–SiC composites. Several industries and institutions are working on C/C–SiC composites to investigate them as friction materials for brake discs and brake pads [11–14]. In the early 1990s, Krenkel [15] investigated C/C–SiC composites for high performance automobile applications. Till now, C/C–SiC composites have been applied to Porsche, Ferrari, and Daimler Chrysler [16]. Various reports on the self-friction of C/C–SiC composites show that friction performance and stability of the composites could be improved with the reduction of porosity and increase in carbon content [17–19].

C/C composites show extremely unsuitable tribological behaviour in adverse environments. Thus C/C composites are rarely used in wet and corrosive environments [20,21]. However, C/C–SiC composites can be used as brake materials in the wet and high speed

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braking conditions [8,9,22].

2 Processing of C/C–SiC composites

The properties of C/C–SiC composites depend on several factors such as type, amount, and microstructure of the matrix and fibres and their interface as well [23]. Thus the properties of C/C–SiC composites greatly depend on the manufacturing methods. The manufacturing methods of carbon fibre reinforced silicon carbide composites mainly differ in the way by which the SiC matrix is built up. Currently, three processes are used for the manufacturing of carbon fibre reinforced SiC matrix composites, i.e., chemical vapour infiltration (CVI), liquid polymer infiltration (LPI) also called polymer infiltration and pyrolysis (PIP), and melt infiltration (MI) or liquid silicon infiltration (LSI). CVI and LPI processes are expensive due to high material costs caused by fibre coatings. CVI and LPI are also time taking processes and their production cycle time can go from several weeks to months. Thus in the late 1980s, cost efficient and less time consuming process was developed, i.e., MI/LSI process, which opens new application fields for carbon fibre reinforced ceramic matrix composites [24].

In MI/LSI process, porous carbon–carbon composites are impregnated by molten silicon, and carbon reacts with silicon to form silicon carbide. Firstly, a CFRP (carbon fibre reinforced plastic) preform is made by common techniques like resin transfer moulding (RTM), autoclave technique, or warm pressing. The use of as received carbon fibres (without any coating) leads to strong fibre–matrix bonding in CFRP composites. The extent of fibre–matrix bonding in CFRP composites is measured in terms of interlaminar shear strength (ILSS) which typically lies in the range of 40–50 MPa for bi-directionally reinforced CFRP composites which are used as preform for LSI process [25]. In the second step, the CFRP composites are pyrolyzed in inert gas atmosphere (N_2) at temperatures between 900 and 1650 ℃, which transforms the polymer matrix into a highly porous C/C preform. As the pyrolysis temperature is increased, shrinkage stress generates which locally exceeds the tensile strength of matrix and leads to cracking. It was observed by using thermo-optical analysis that in the case of bi-directionally reinforced CFRP composites, first fibre–matrix debonding occurs beyond 505 ℃ [26]. This debonding results in the segmentation of the fibre tows and leads to a translaminar microcrack pattern with dense C/C segments [25]. In the last step, the fibre preform having porosity is infiltrated by molten silicon at temperature beyond the melting point of silicon, i.e., 1415 ℃, mostly applying only capillary forces. Typically the operating temperature is in the range of 1600 ℃.

Critical parameters which are to be considered in the processing of C/C–SiC composites by LSI technique are low melt viscosity, good wetting of fibre reinforcement, high chemical reactivity, and anomaly of silicon during the phase transition (density change of silicon is approximately 8%).

After infiltration, the molten silicon fills the cracks within some minutes. The dense carbon matrix shields the fibre bundles from highly reactive silicon, and a layer of silicon carbide is formed around the segments. Thus only a small amount of carbon fibres are converted and damaged. If carbon fibres are thermally treated before implementation in the polymer matrix, the active surface groups are de-activated which leads to reduced fibre–matrix bonding. Thus pyrolysis results in carbon fibre tows without any segmentation, and liquid silicon can react with and damage most of the carbon fibres. Thus load bearing capacity of the composites is reduced and fracture behaviour of the composites is similar to monolithic ceramics [27,28].

Many manufacturers have developed different C/C–SiC composites which are characterised by low porosity (lower than 6%), high fibre volume content (higher than 30%), and moderate mechanical properties. Table 1 gives an overview of the properties of C/C–SiC brake materials with a 2D fibre reinforcement.

Table 1 C/C–SiC brake materials with a 2D fibre reinforcement[a]

	C/C–SiC (2D reinforcement)		
	Schunk CF226/2 P77	DLR silca XS	DLR silca XG
Density (g/cm^3)	1.95	1.8–1.9	2.1–2.3
Flexural strength (MPa)	130	> 160	80
Young's modulus (GPa)	55	63	41
Open porosity (%)	< 6	< 3	< 5
CTE_{II} ($10^{-6}K^{-1}$)	—	0.4–2.6[b]	1.1–2.5[c]
Thermal conductivity$_I$ (W/mK)	20	15.3	18.9
Thermal conductivity$_{II}$ (W/mK)	30	26.3	—
SiC content (%)	25	30	50[c]
Process	LSI	LSI	LSI

[a] Reproduced with permission from Ref. [25], © 2005 Elsevier B.V.
[b] –50 up to 1600 ℃.
[c] Si plus SiC content.

3 Tribological behaviour of C/C–SiC composites

Friction brakes decelerate a vehicle by transforming the kinetic energy of the vehicle into heat and dissipate that heat into surroundings. The brakes which are used in automotive braking system are generally made of steel or grey cast iron and coupled with polymer composite pads [29]. These materials are used in braking system with moderate loads and limited capability to withstand temperature. These materials have relatively high and stable coefficient of friction and low wear rate, and are quiet during operation while used in light vehicles [30].

However, heavy vehicles require more power for braking than conventional braking system. To accommodate that, C/C composites [31,32] came into picture which can provide more friction and operate at high temperature. These types of high speed brakes require 4500 J/cm^2 of kinetic energy per unit area of sliding surfaces to be converted into heat in approximately 30 seconds. The thermal cycling leads to cracking of uncoated C/C composites, which leads to accelerated wear and oxidation [33].

The main requirements of high performance brake materials are:

- stable coefficient of friction (dynamic and static);
- low wear rate for increased life;
- low life cycle cost;
- low weight;
- high degree of freedom in the structural design (for internal cooling ducts, attachments).

Therefore, in recent years, C/C–SiC composites were developed as high performance brake materials following metallic materials and C/C composites [34].

The presence of SiC in the composite matrix results in increase in coefficient of friction, but wear rate also increases due to abrasive action of silicon carbide [9]. SiC generally damages by fatigue [35]. Carbon fibres get loose by repeated flexion due to debonding of SiC grains. It was observed that the effect of SiC is less when it is deposited by CVI [36]. The presence of free Si in the composite matrix has a great influence on the third body morphology. Adhesive friction also comes into play due to the presence of free silicon, and there is an increase in coefficient of friction and a decrease in wear. The adhesive component of friction increases due to presence of free silicon. Silicon is less hard than silicon carbide, due to which free silicon is subjected to plastic deformation which increases adhesion [9].

Figure 1 shows emergency braking system and internally ventilated brake disc for passenger cars made of C/C–SiC composites. There should be a flow of heat from outer region to the centre of the composites to prevent the overheating of the friction surface (outer region). Thus C/C–SiC composites should have high transverse thermal conductivity. To meet the essential requirement, C/C–SiC composites should be modified in their composition and microstructure. Increase in silicon carbide content of the composites leads to increase in the transverse thermal conductivity of the C/C–SiC composites [8]. However, low carbon content and increase in silicon carbide content lead to decrease in damage tolerance of the composites and also influence the mechanical properties of the composites. High SiC content is mainly required in the outer region of the disc. Two approaches have been suggested to fulfil the requirement of high silicon carbide content in the outer region [25] as follows:

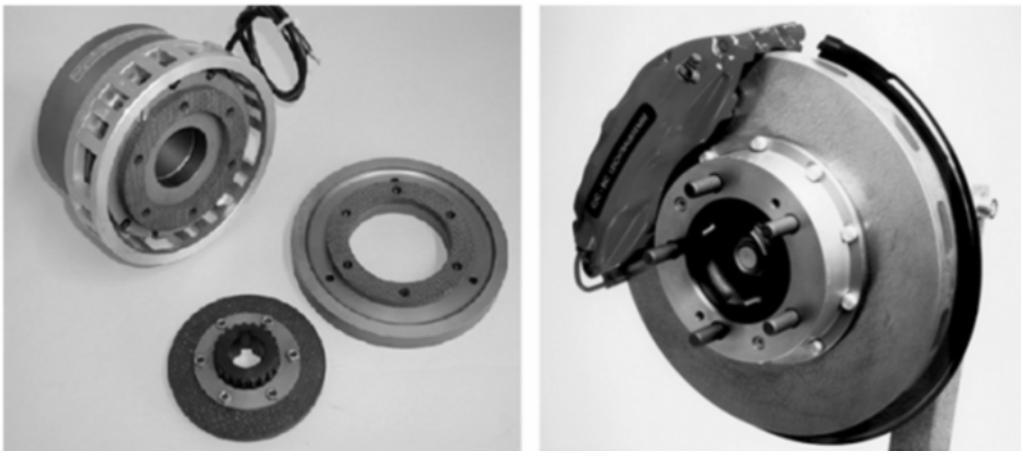


Fig. 1 Emergency braking system (left) and internally ventilated brake disc for passenger cars (right), made of C/C–SiC composites. Reproduced with permission from Ref. [37], © 2005 Kluwer Academic Publishers.

- gradual increase of SiC from the centre to the outer region (i.e., friction surface);
- homogeneous C/C–SiC composites with Si–SiC coating on the outer region.

These two approaches successfully meet the requirement, but due to higher coefficient of thermal expansion of ceramic rich surface than that of C/C–SiC substrate, tensile stress develops in the surface during cooling after processing leading to micro cracking of the surface. Braking heats up the surface. Thus cracks get narrower as outer region expands more than substrate. Normally the cracks formed run throughout the thickness but stop in the ductile core region. Thus the breakage of fibres does not occur.

In disc brakes, the material of pad also influences the tribological properties of C/C–SiC composites. The composition of pad material influences the coefficient of friction. Even slight change in the composition of pad material can change the coefficient of friction. Results of most of the tests and pad material's composition are confidential and not published. Some results are shown in Figs. 2 and 3.

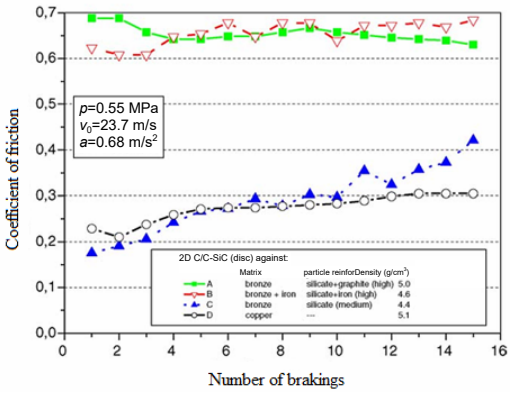


Fig. 2 Influence of the pad material on the tribological behaviour of 2D reinforced C/C–SiC brake discs. Reproduced with permission from Ref. [25], © 2005 Elsevier B.V.

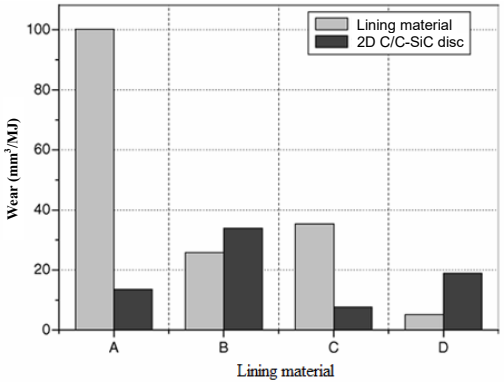


Fig. 3 Wear rate of C/C–SiC brake discs in combination with commercial lining materials (for material's composition see Fig. 2). Reproduced with permission from Ref. [25], © 2005 Elsevier B.V.

There are so many factors which affect the tribological performance of C/C–SiC composites. Some of them are described here.

3.1 Effect of pyrolytic carbon content on tribological properties of C/C–SiC composites

The amount of pyrolytic carbon in matrix has great effect on its final performance because C/C–SiC composites are manufactured from porous C/C preform.

The variation of tribological properties with the weight percentage of pyrolytic carbon is shown in Fig. 4. In Fig. 4, it is shown that coefficient of friction increases as weight percentage of pyrolytic carbon increases up to some extent, and then coefficient of friction decreases. Braking time decreases very rapidly with increase in

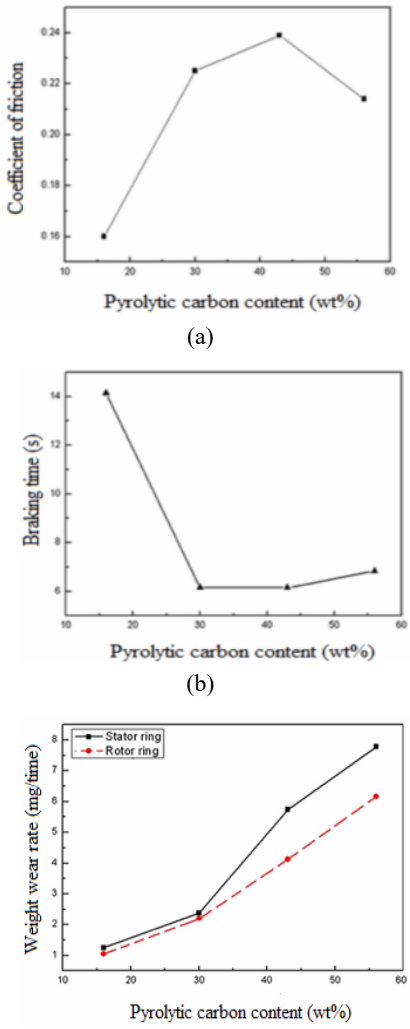


Fig. 4 Dependence of tribological properties of C/C–SiC composites on the content of pyrolytic carbon in matrix: (a) average coefficient of friction, (b) braking time, and (c) weight wear rate. Reproduced with permission from Ref. [38], © 2011 Elsevier Masson SAS.

pyrolytic carbon content, but after some extent it becomes almost constant.

During braking, superficial hardness can increase the coefficient of friction by enhancing the ploughing effect [19]. Superficial hardness decreases with increase in pyrolytic carbon content which results in increase in coefficient of friction, because friction surface gets softer and hard SiC asperities may pierce with ease [38].

3. 2 Effect of porosity and SiC content on tribological properties of C/C–SiC composites

C/C–SiC composites can be fabricated by various methods, i.e., CVI, PIP, LSI, and so on [39,40]. Each fabrication method has its advantages and drawbacks. The densification is very slow in the cases of CVI and PIP. Materials fabricated by CVI and PIP contain some residual porosity. However, materials fabricated by LSI process display poor mechanical properties [40] and contain some free silicon which affects the tribological properties [17].

Porosity and SiC content decrease with increase in the initial preform density. Figure 5 shows the curves of coefficient of friction for samples from different methods. As braking starts, there is a sudden pressure on the surface of brakes, and the first part of the curves in Fig. 5 is fluctuating. However, due to embedment of grains, there generate new abrasive particles which lead to rapid increase of friction [1]. The coefficient of friction may reduce due to pulverization of some abrasive particles which act as friction film. There is a competition between abrasive action and lubrication action which tends to change the curves (Fig. 5) smoothly. Temperature rise in the material leads to the decomposition of the material, which leads to increase in the coefficient of friction [1,18].

High rigidity, abrasion resistance, corrosion resistance, and high thermal conductivity of SiC have a great influence on the tribological properties of C/C–SiC composites. Due to abrasive nature of SiC particles, it enhances the coefficient of friction. However, when the content of SiC is very high, it acts as lubricant and may reduce the coefficient of friction [41,42]. High porosity may stunt the formation of SiC framework and spur the formation of abrasive film which results in the deterioration of tribological properties [1,41] (results can be analysed from Figs. 6 and 7 through comparison). Figure 6 shows the porosity as the weight percentage of silicon carbide varies.

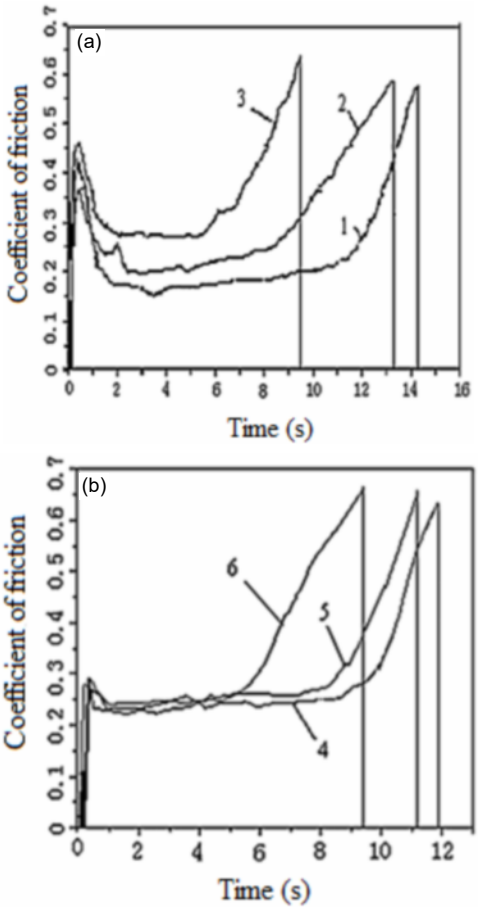


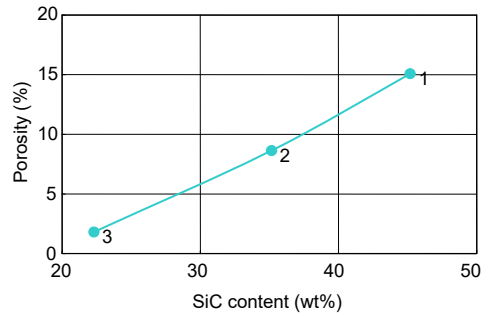
Fig. 5 Curves of coefficient of friction for samples from two different methods: (a) PIP and (b) CVI. Details of samples 1, 2, 3, 4, 5, and 6 are given in Fig. 6. Reproduced with permission from Ref. [1], © 2011 Elsevier Ltd and Techna Group S.r.l.

Figure 7 shows the variation of coefficient of friction with the variation of weight percentage of silicon. Coefficient of friction decreases with increase in SiC content for both fabrication techniques. However, coefficient of friction is more in the case of CVI fabrication technique than PIP fabrication technique.

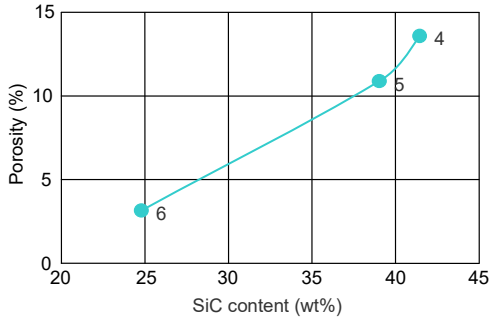
C/C–SiC composites have high porosity due to which it possess high wear rate. This is attributed to the fact that due to higher porosity, SiC particle framework does not easily form and the hard SiC particles plough and micro-cut the friction surface resulting in grain abrasion [17].

3. 3 Effect of braking pressure and braking speed on tribological properties

In brake disc, kinetic energy is converted into heat at the time of braking. Braking energy increases as the speed of braking increases.

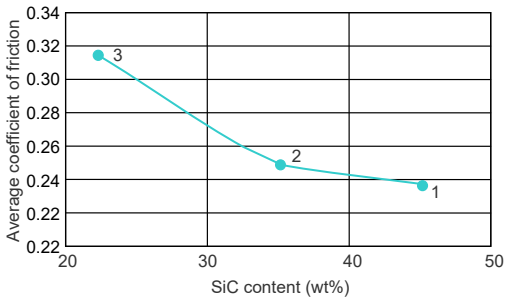


(a) PIP

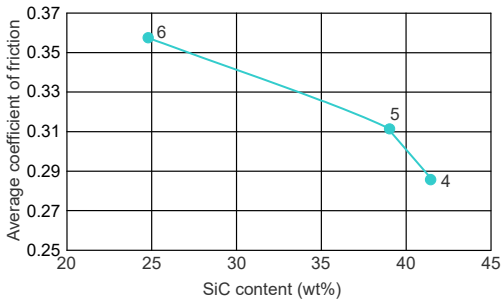


(b) CVI

Fig. 6 Porosity as a function of SiC content: (a) PIP and (b) CVI. Reproduced with permission from Ref. [1], © 2011 Elsevier Ltd and Techna Group S.r.l.



(a) PIP



(b) CVI

Fig. 7 Average coefficient of friction as a function of SiC content: (a) PIP and (b) CVI. Reproduced with permission from Ref. [1], © 2011 Elsevier Ltd and Techna Group S.r.l.

Figure 8 shows the effect of braking pressure and braking speed on the average coefficient of friction of C/SiC brake materials. The carbon fibre, pyrolytic carbon, and SiC are brittle in nature. At low braking speed, the impact and shear force between the asperities are small and the asperities cannot be completely sheared [43]. However, when braking speed increases, braking energy increases and so many asperities create brittle fracture. So due to brittle fracture, there is a formation of many new asperities and little debris, which results in increase in the coefficient of friction. When the braking speed is high enough to cut the asperities, many asperities are sheared and pulverized by impact and shear action between asperities, and a lot of debris forms. The debris formed fills the gap between peaks and valleys which decreases the ploughing effect. Due to increase in braking speed, more and more debris forms, which forms a continuous film on the surface resulting in decrease of coefficient of friction and friction resistance. Figure 9 shows the schematic illustration of the debris filled and the friction film on the surface [44].

When braking speed increases, wear rate increases due to increase in impact and shear force between asperities and ploughing action between asperities. Temperature also increases with increase in braking speed which leads to the oxidation of carbon matrix and carbon fibres near the surface. Thus wear rate increases with increase in pressure.

Figure 10 shows the effect of braking pressure and braking speed on the wear rate of C/SiC brake materials. It can be seen that at high braking speed, wear rate increases rapidly. At low braking speed, the effect of increase in pressure on wear rate is not significant. At high braking speed, increase in pressure leads to increase in wear rate.

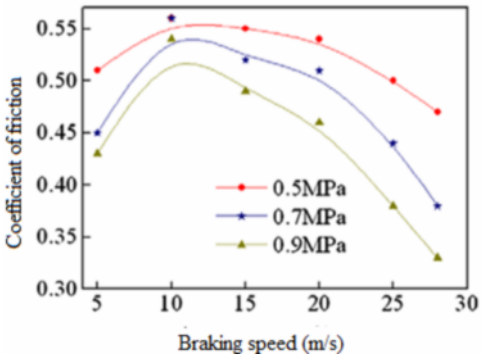
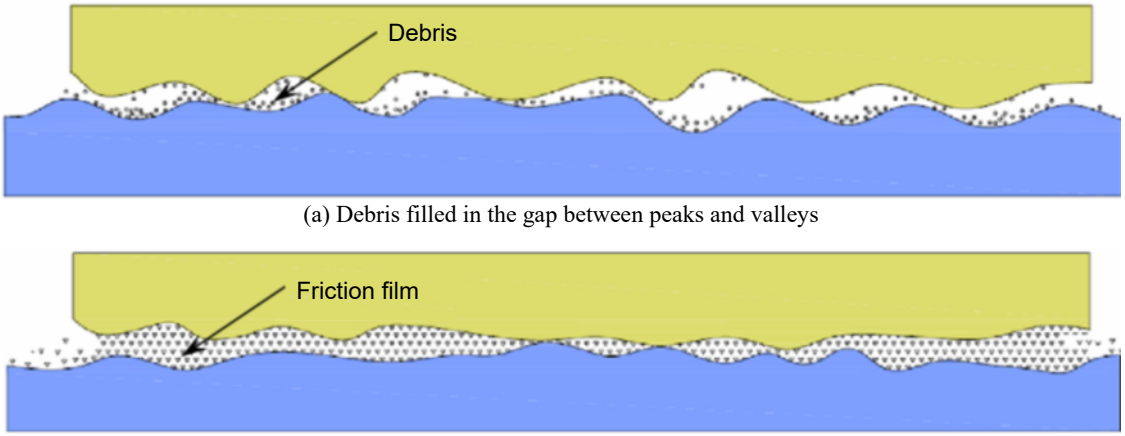


Fig. 8 Effect of braking pressure and braking speed on the average coefficient of friction of C/SiC brake materials. Reproduced with permission from Ref. [43], © 2010 Elsevier Ltd.



(a) Debris filled in the gap between peaks and valleys

(b) Friction film formed between the friction surfaces

Fig. 9 Schematic illustration of the debris filled and the friction film formed. Reproduced with permission from Ref. [44], © 2001 Elsevier Inc.

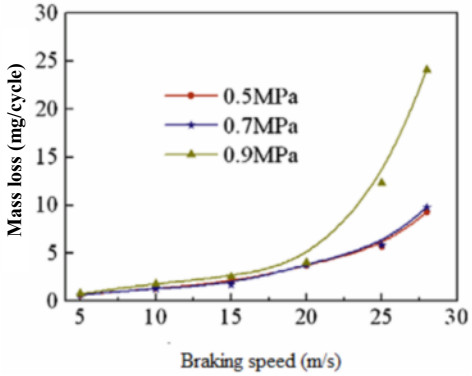


Fig. 10 Effect of braking pressure and braking speed on the wear rate of C/SiC brake materials. Reproduced with permission from Ref. [43], © 2010 Elsevier Ltd.

3. 4 Tribological performance under dry and wet conditions

The tribological behaviour of C/C–SiC composites under dry and wet conditions is different. Most of the studies on tribological behaviour of C/C–SiC composites focus on dry condition [40,41,45]. The studies on wet condition are less. But tribological behaviour under wet condition is a key indicator under braking conditions [8,40,46]. The characteristics of C/C–SiC braking composites change with environmental conditions, which is one of the great advantage as compared to other braking composites [47]. The coefficient of friction does not change much with the condition (either dry or wet), but change in wear rate with condition is significant. The variation of coefficient of friction under dry and wet conditions is shown in Fig. 11.

Plastic deformation and abrasion caused by ploughing are the main mechanisms of wear of C/C–SiC composites under dry condition. This results in high coefficient of friction and wear rate as compared to wet condition. Table 2 shows the tribological performance of C/C–SiC under dry and wet conditions.

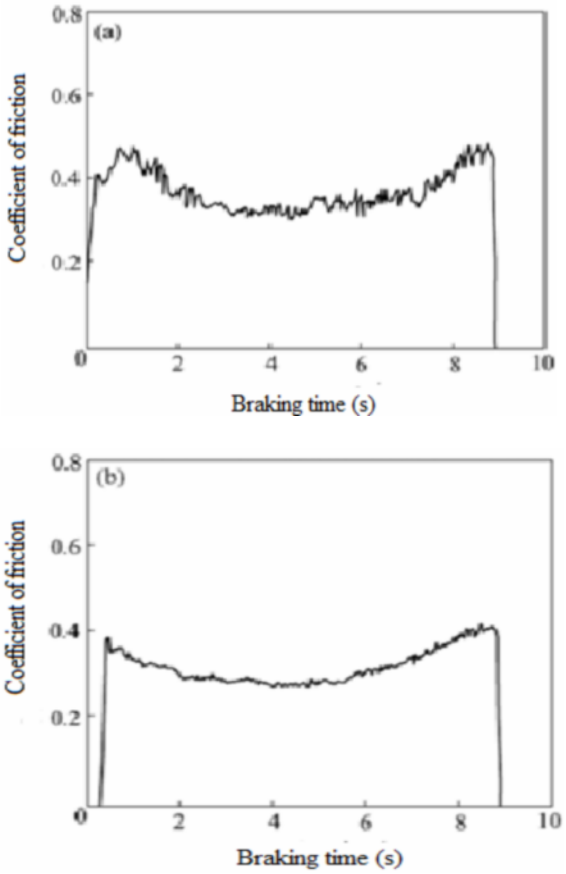


Fig. 11 First braking curves of C/C–SiC on different conditions: (a) dry condition and (b) wet condition. Reproduced with permission from Ref. [46], © 2008 The Nonferrous Metals Society of China.

Table 2 Tribological performance of C/C–SiC under dry and wet conditions [47]

Braking condition	Coefficient of friction	Stability coefficient	Braking power (W/cm^2)	Braking energy (J/cm^2)	Thickness loss (μm)	Counterpart thickness loss (μm)
Dry	0.38	0.69	193.40	1245.36	1.10	0.97
Wet	0.35	0.64	196.72	1242.78	0.70	0.63

However, there are sticking sliding and plastic deformation under wet condition. That is why the coefficient of friction and wear loss are less in wet condition as compared to dry condition.

C/C–SiC composites sliding under dry condition contain hard phase SiC, resin carbon, and second hard phase Si (silicon) particles. Hard phase Si particles cannot be cut by the shearing and compressive force and can easily be broken due to their lower plasticity. The coefficient of friction increases due to these hard particles at the early stage.

Under wet condition, there is a presence of moisture. Due to frictional heat, the moisture at the surface evaporates and reacts with SiC matrix and forms $SiO_2 \cdot xH_2O$ or H_2SiO_3 [48,49]. This results in formation of friction films. These friction films combine with the subsurface of C/C–SiC composites, which leads to low shearing strength and high brittleness of the composites. They tend to break partially due to the existence of shearing and compressive force [50]. The worn debris unites and tries to fill the pits in the worn surface. The reunited debris spreads continually unless the formation of friction films takes place again. This results in decrease in coefficient of friction and wear rate. The main wear mechanisms under wet condition are oxidation and adhesion.

3. 5 Effect of graphitization of C/C preform on tribological properties of C/C–SiC composites

The processing of porous C/C preform is a very important step because the processing parameters affect the microstructure and properties of the preform. Tribological properties depend upon the microstructure and properties of the preform [8,51].

The effect of graphitization on microstructure and mechanical properties of C/C–SiC composites is investigated by many researchers, but its effect on tribological properties of C/C–SiC composites has not been investigated much [52–56].

Figure 12 shows the curves of coefficient of friction of samples ((a) without graphitization and (b) with graphitization) as a function of braking time. It is clear from Fig. 12 that at the start of braking, coefficient of friction is high for composites without graphitization than composites with graphitization. This leads to locking of braking system. Thus it is harmful for braking. This locking phenomenon can be greatly modified by graphitization of C/C preform. The nature behind this phenomenon is explained earlier in Section 3.2.

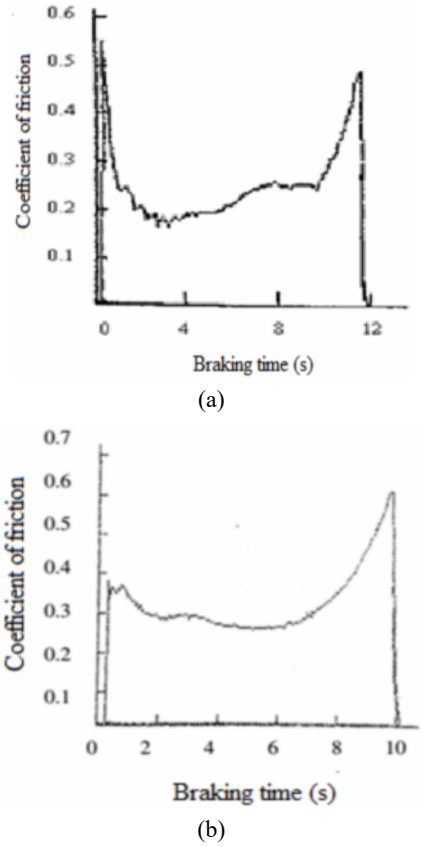


Fig. 12 Curves of coefficient of friction of samples as a function of braking time: (a) without graphitization and (b) with graphitization. Reproduced with permission from Ref. [18], © 2008 Elsevier Ltd.

Hardness of the material plays an important role in tribology [57,58]. Graphitization increases the content of hard SiC in C/C–SiC composites. Due to graphitization, some graphite crystallite forms [51], which results in softening of turbostratic structure carbon. During braking, hard SiC inserts into soft turbostratic structure carbon easily and deeply [53]. This results in increase in coefficient of friction. However, graphite has a good solid lubricant property because it has weakly bonded layer structure [53,56]. This lubricating property helps to avoid the initial high peak in coefficient of friction and moderate the fluctuation in coefficient of friction in the latter stage. In spite of this lubricating effect, the coefficient of friction

of graphitized C/C–SiC composites is higher in the latter stage. This cannot be explained only by lubricating property. In the latter stage, the effect of material hardness is more than lubricating action. Figure 13 shows the coefficient of friction as a function of number of braking tests.

The wear rate of graphitized C/C–SiC composites is higher than that of non-graphitized C/C–SiC composites. This is due to shearing of large amount of soft graphite particles by ploughing effect under high braking load. Though wear rate of graphitized C/C–SiC composites is higher, it is under acceptable limits.

3. 6 Effect of C/C composites and C/C–SiC composites as counterface material

C/C composites possess high strength, high thermal conductivity, high heat capacity, low coefficient of thermal expansion, good wear property, low density, and self-lubricating capability [59,60]. However, C/C composite brakes have poor oxidation resistance and unstable coefficient of friction [61–64]. The static coefficient of friction of C/C composites is very low. To overcome these problems, C/C–SiC composites are developed. C/C–SiC composites have stable tribological properties, low wear rate, and long life [8,65]. C/C–SiC composites have good resistance to oxidation and high static coefficient of friction, and the sensibility to environment is low. This makes C/C–SiC composites a potential material for aircraft braking system [11,25,28,66,67]. However, the density of C/C–SiC composites is higher than C/C composites which makes C/C–SiC composites heavier than C/C composites and limits their applications in some weight sensitive aircrafts.

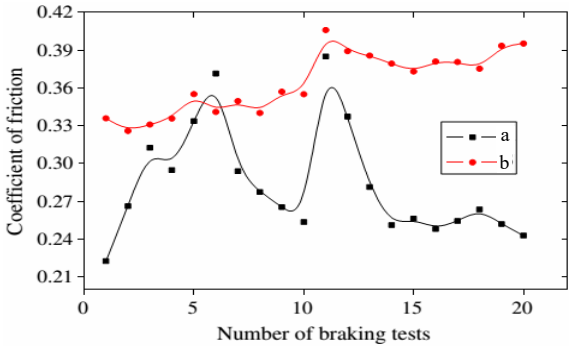
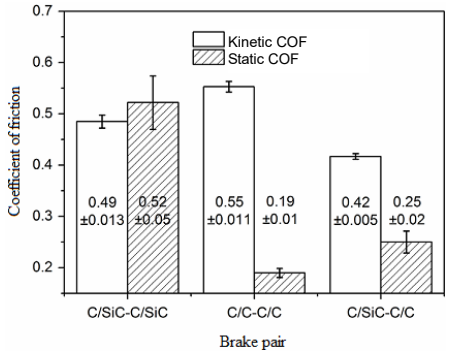
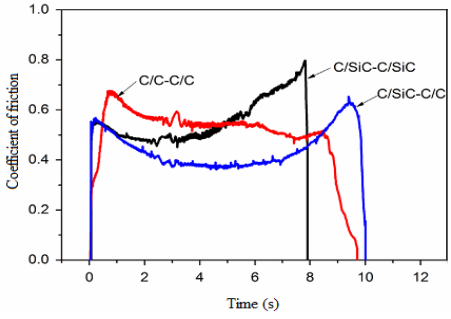


Fig. 13 Coefficient of friction as a function of number of braking tests: (a) without graphitization and (b) with graphitization. Reproduced with permission from Ref. [18], © 2008 Elsevier Ltd.

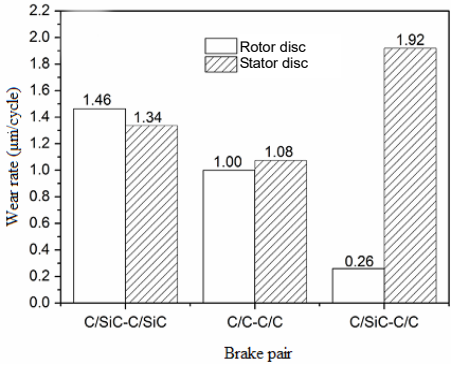
Some matrix modification can be done to overcome the disadvantages of C/C and C/C–SiC composites [62,68–71], but this requires a lot of time and plenty of experiments. Thus to meet the requirements of high performance and light weight for weight sensitive aircrafts, C/C–SiC and C/C brake pair can be used instead of using the same material on both discs [72]. Figure 14 shows the tribological and wear properties of different brake pair systems in dry condition. Figure 14 indicates the comparative behaviours of C/C and C/C–SiC brake pair, C/C disc as the stator disc, and C/C– SiC disc as the rotor disc.



(a)



(b)



(c)

Fig. 14 Tribological and wear properties in dry condition: (a) kinetic and static coefficients of friction (COF), (b) typical brake curves, and (c) linear wear rates. Reproduced with permission from Ref. [72], © 2014 Elsevier Ltd.

The kinetic coefficient of friction of C/C–SiC and C/C brake pair is little lower than that of C/C self mated pair, but the static coefficient of friction is higher. Thus taxiing distance of aircrafts before take-off can be reduced [73]. At the end of braking, the coefficient of friction of C/C and C/C–SiC brake pair is higher than that of C/C self mated pair, due to which braking distance reduces. As more wear occurs in C/C brake disc than in C/C–SiC brake disc, C/C disc is made thicker and C/C–SiC disc is made thinner, which also reduces the weight of aircraft brake assembly. Due to high heat capacity of C/C composites [60], the C/C and C/C–SiC brake pair does not suffer much from high surface temperature.

The nature of braking curve of C/C–SiC self mated braking pair can be explained by shearing, deformation, breaking of asperities, ploughing effect, etc., as explained earlier in this paper. However, in the cases of C/C self mated and C/C–SiC brake pairs, C/C composites easily form flaky debris, and flaky debris easily forms friction films. Due to the formation of friction films, coefficient of friction reduces and gets stabilized. In the case of C/C and C/C–SiC brake pair, ploughing effect increases the coefficient of friction and wear rate of C/C disc.

4 Summary

Due to their low density, high thermal shock resistance, and good abrasive resistance, C/C–SiC composites can be used for clutch and braking systems. C/C–SiC composites show better tribological properties than grey cast iron and C/C composites. The tribological properties of C/C–SiC composites depend on many parameters. However, the nature of braking curve remains almost the same but the value varies. In brake discs, the pad material and composition greatly influence the tribological properties of C/C–SiC composite brakes. And in clutches, the material of mating disc greatly influences the tribological properties of C/C–SiC discs. Wear resistance can be highly improved by gradual increase of SiC from the centre to outer region or by using Si–SiC coatings on the outer surface. Researchers have tried to tailor the microstructure and properties of C/C–SiC composites. Many of them have succeeded to a greater extent. But there is always a chance for improvement. Now the main focus is on quality assurance and reduction in material cost of C/C–SiC composites.

References

[1] Zhou X, Zhu D, Xie Q, *et al*. Friction and wear properties of C/C–SiC braking composites. *Ceram Int* 2012, **38**: 2467–2473.

[2] Cui P, Chen Z, Li S. Rapid densification and properties of C/C brake discs. *Acta Materiae Compositae Sinica* 2008, **25**: 101–105.

[3] Tai NH, Kuo HH, Chern Lin JH, *et al*. Mechanical and tribological properties of 2-D carbon/carbon composites densified through pulse chemical vapor infiltration. *J Mater Sci* 2002, **37**: 3693–3703.

[4] Kuo HH, Chern Lin JH, Ju CP. Effect of a post-treatment on tribological behaviour of PAN/CVI and pitch/phenolic/CVI-based C/C composites. *J Mater Sci* 2005, **40**: 3263–3265.

[5] Liao X, Li H, Xu W, *et al*. Study on the thermal expansion properties of C/C composites. *J Mater Sci* 2007, **42**: 3435–3439.

[6] Wang D, Liu Y. Present situation of friction materials. *Adv Ceram* 2007, **3**: 15–19.

[7] Hokao M, Hironaka S, Suda Y, *et al*. Friction and wear properties of graphite/glassy carbon composites. *Wear* 2000, **237**: 54–62.

[8] Krenkel W, Heidenreich B, Renz R. C/C–SiC composites for advanced friction systems. *Adv Eng Mater* 2002, **4**: 427–436.

[9] Fouquet S, Rollin M, Pailler R, *et al*. Tribological behaviour of composites made of carbon fibres and ceramic matrix in the Si–C system. *Wear* 2008, **264**: 850–856.

[10] Li Z, Xiao P, Xiong X, *et al*. Preparation and tribological properties of C fibre reinforced C/SiC dual matrix composites fabrication by liquid silicon infiltration. *Solid State Sci* 2013, **16**: 6–12.

[11] Krenkel W, Henke T. Design of high performance CMC brake discs. *Key Eng Mat* 1999, **164–165**: 421–424.

[12] Li Z, Xiao P, Xiong X. Preparation and properties of C/C–SiC brake composites fabricated by warm compacted-*in situ* reaction. *Int J Min Met Mater* 2010, **17**: 500–505.

[13] Gadow R, Kienzle A. Processing and manufacturing of C-fibre reinforced SiC-composites for disk brakes. In: Proceedings of the 6th International Symposium on Ceramic Materials and Components for Engines, 1997: 412–418.

[14] Xiao P, Li Z, Xiong X. Microstructure and tribological properties of 3D needle punched C/C–SiC brake composites. *Solid State Sci* 2010, **12**: 617–623.

[15] Krenkel W. CMC materials for high performance brakes. In: Proceedings of ISATA Conference on Supercars, 1994: 769–775.

[16] Fan S, Zhang L, Xu Y, *et al*. Microstructure and tribological properties of advanced carbon/silicon carbide aircraft brake materials. *Compos Sci Technol* 2008, **68**: 3002–3009.

[17] Fan S, Zhang L, Cheng L, *et al*. Wear mechanisms of the C/SiC brake materials. *Tribol Int* 2011, **44**: 25–28.

[18] Jiang G, Yang J, Xu Y, *et al*. Effect of graphitization on microstructure and tribological properties of C/SiC

composites prepared by reactive melt infiltration. *Compos Sci Technol* 2008, **68**: 2468–2473.

[19] Xu Y, Zhang Y, Cheng L, *et al*. Preparation and friction behaviour of carbon fibre reinforced silicon carbide matrix composites. *Ceram Int* 2007, **33**: 439–445.

[20] Chen JD, Chern Lin JH, Ju CP. Effect of humidity on the tribological behaviour of carbon–carbon composites. *Wear* 1996, **193**: 38–47.

[21] Shim HH, Kwon OK. Effects of fibre orientation and humidity on friction and wear properties of graphite fibre composites. *Wear* 1992, **157**: 141–149.

[22] Fan S, Zhang L, Xu Y, *et al*. Microstructure and properties of 3D needle-punched carbon/silicon carbide brake materials. *Compos Sci Technol* 2007, **67**: 2390–2398.

[23] Zhang Y, Xiao Z, Wang J, *et al*. Effect of pyrocarbon content in C/C preforms on microstructure and mechanical properties of the C/C–SiC composites. *Mat Sci Eng A* 2009, **502**: 64–69.

[24] Krenkel W, Hald H. Liquid infiltrated C/SiC—An alternative material for hot space structures. In: Proceedings of the ESA/ESTEC Conference on Spacecraft Structures and Mechanical Testing, 1988.

[25] Krenkel W, Berndt F. C/C–SiC composites for space applications and advanced friction systems. *Mat Sci Eng A* 2005, **412**: 177–181.

[26] Schulte-Fischedick J, Frieß M, Krenkel W, *et al*. Crack microstructures during the carbonizing of carbon fiber reinforced plastics to carbon/carbon composites. In: Proceedings of the 12th International Conference on Composite Materials, 1999.

[27] Krenkel W. Designing with C/C–SiC composites. In: *Advances in Ceramic Matrix Composites IX, Volume 153*. Bansal NP, Singh JP, Kriven WM, *et al*. Eds. Hoboken, NJ, USA: John Wiley & Sons, Inc., 2006: 103–123.

[28] Krenkel W. C/C–SiC composites for hot structures and advanced friction systems. In: Proceedings of the 27th Annual Cocoa Beach Conference on Advanced Ceramics and Composites: B: Ceramic Engineering and Science. Kriven WM, Lin H-T, Eds. Hoboken, NJ, USA: John Wiley & Sons, Inc., 2003, **24**: 583–592.

[29] Jacko MG, Tsang PHS, Rhee SK. Automotive friction materials evolution during the past decade. *Wear* 1984, **100**: 503–515.

[30] Rhee SK, Jacko MG, Tsang PHS. The role of friction film in friction wear and noise of automotive brakes. *Wear* 1991, **146**: 89–97.

[31] Daimler Chrysler AG. Ceramic brake discs for high performance vehicles. Daimler Chrysler High Tech Report 96, 2000.

[32] Shin H-K, Lee H-B, Kim K-S. Tribological properties of pitch-based 2-D carbon–carbon composites. *Carbon* 2001, **39**: 959–970.

[33] Yen BK, Ishihara T. On temperature-dependent tribological regimes and oxidation of carbon–carbon composites up to 1800 ℃. *Wear* 1996, **196**: 254–262.

[34] Krenkel W. Design of ceramic brake pads and disks. In: Proceedings of the 26th Annual Conference on Composites, Advanced Ceramics, Materials, and Structures: A: Ceramic Engineering and Science. Lin H-T, Singh M, Eds. Hoboken, NJ, USA: John Wiley & Sons, Inc., 2002, **23**: 319–329.

[35] Fournier P, Reynaud Ph, Platon F, *et al*. Tribological behaviour of carbon-fibre-reinforced SiC matrix composites. *Proc. IMechE Part J: J Engineering Tribology* 2000, **214**: 291–306.

[36] Fillion A. Composites C/C et C/C–SiC pour applications tribologiques. Ph.D. thesis from Bordeaux I University, France, No. 2168, 2000.

[37] Krenkel W. Carbon fibre reinforced silicon carbide composites (C/SiC, C/C–SiC). In: *Handbook of Ceramic Composites*. Bansal NP, Ed. Kluwer Academic Publishers, 2005: 117–148.

[38] Shi Q, Xiao P. Effect of pyrolytic carbon content on microstructure and tribological properties of C/C–SiC brake composites fabricated by isothermal chemical vapor infiltration. *Solid State Sci* 2012, **14**: 26–34.

[39] Zhang J, Fan S, Zhang L, *et al*. Microstructure and frictional properties of 3D needled C/SiC brake materials modified with graphite. *T Nonferr Metal Soc* 2010, **20**: 2289–2293.

[40] Xiao P, Xiong X, Zhang H, *et al*. Progress and application of C/C–SiC ceramic braking materials. *The Chinese Journal of Nonferrous Metals* 2005, **15**: 667–674.

[41] Xiao P, Xiong X, Ren Y. Effect and mechanism of different components of C/C–SiC composites on friction and wear behaviours. *The Chinese Journal of Nonferrous Metals* 2005, **15**: 1040–1044.

[42] Zhang Y, Xu Y, Lou J, *et al*. The analysis of friction and wear performance of C/C–SiC composites. *Journal of Aeronautical Materials* 2005, **25**: 49–54.

[43] Fan S, Zhang L, Cheng L, *et al.* Effect of braking pressure and braking speed on the tribological properties of C/SiC aircraft brake materials. *Compos Sci Technol* 2010, **70**: 959–965.

[44] Stachowiak GW, Batchelor AW. *Engineering Tribology*, 2nd edn. Butterworth-Heinemann, 2001.

[45] Heidenreich B, Renz R, Krenkel W. Short fibre reinforced CMC materials for high performance brakes. In: Proceedings of the 4th International Conference on High Temperature Ceramic Matrix Composites, 2001: 68–74.

[46] Li Z, Xiao P, Xiong X, *et al*. Tribological characteristics of C/C–SiC braking composites under dry and wet conditions. *T Nonferr Metal Soc* 2008, **18**: 1071–1075.

[47] Xiong X, Huang B, Xu H, *et al*. Frictional and wear behaviours of C/C composites from carbon fibre cloth at different braking speeds. *The Chinese Journal of Nonferrous Metals* 2002, **12**: 255–259.

[48] Anderson P. Water-lubricated pin-on-disk tests with ceramics. *Wear* 1992, **154**: 37–47.

[49] Lancaster JK. A review on the influence of environmental humidity and water on friction lubrication and wear. *Tribol Int* 1990, **23**: 371–389.

[50] Zum Gahr K-H. Sliding wear of ceramic–ceramic, ceramic–steel and steel–steel pairs in lubricated and unlubricated contact. *Wear* 1989, **133**: 1–22.

[51] Tzeng S-S, Chr Y-G. Evolution of microstructure and properties of phenolic resin-based carbon/carbon

composites during pyrolysis. *Mater Chem Phys* 2002, **73**: 162–169.
[52] Frieß M, Krenkel W. Influence of the graphitization temperature on the properties of C/C–SiC manufactured via LSI-processing. In: Proceedings of the 9th CIMTEC-World Ceramics Congress & Forum on New Materials, 1998.
[53] Paris J-Y, Vincent L, Denape J. High-speed tribological behaviour of a carbon/silicon carbide composite. *Compos Sci Technol* 2001, **61**: 417–423.
[54] Li ZQ, Lu CJ, Xia ZP, *et al*. X-ray diffraction patterns of graphite and turbostratic carbon. *Carbon* 2007, **45**: 1686–1695.
[55] Kermc M, Kalina M, Vižintin J. Development and use of an apparatus for tribological evaluation of ceramic-based brake materials. *Wear* 2005, **259**: 1079–1087.
[56] Zhou Y, Hirao K, Yamauchi Y, *et al*. Tribological properties of silicon carbide and silicon carbide–graphite composite ceramics in sliding contact. *J Am Ceram Soc* 2003, **86**: 991–1002.
[57] Fischer TE, Zhu Z, Kim H, *et al*. Genesis and role of wear debris in sliding wear of ceramics. *Wear* 2000, **245**: 53–60.
[58] Takadoum J, Zsiga Z, Roques-Carmes C. Wear mechanism of silicon carbide: New observations. *Wear* 1994, **174**: 239–242.
[59] Shin H-K, Lee H-B, Kim K-S. Tribological properties of pitch-based 2-D carbon–carbon composites. *Carbon* 2001, **39**: 959–970.
[60] Cai Y, Yin X, Fan S, *et al*. Tribological behaviour of three dimensional needled ceramic modified carbon/carbon composites in seawater conditions. *Compos Sci Technol* 2013, **87**: 50–57.
[61] Chen JD, Chern Lin JH, Ju CP. Effect of humidity on the frictional behaviour of carbon–carbon composites. *Wear* 1996, **193**: 38–47.
[62] Yen BK. Influence of water vapor and oxygen on the tribology of carbon materials with sp^2 valence configuration. *Wear* 1996, **192**: 208–215.
[63] Blanco C, Bermejo J, Marsh H, *et al*. Chemical and physical properties of carbon as related to brake performance. *Wear* 1997, **213**: 1–12.
[64] Byrne C, Wang Z. Influence of thermal properties on friction performance of carbon composites. *Carbon* 2001, **39**: 1789–1801.
[65] Breuer B, Dausend U. *Advanced Brake Technology*. Kabdwal Book International, 2003: 37–50.
[66] Mühlratzer A, Leuchs M. Applications of non-oxide CMCs. In: *High Temperature Ceramic Matrix Composites*. Krenkel W, Naslain R, Schneider H, Eds. Wiley-VCH Verlag GmbH & Co. KGaA, Weinheim, FRG, 2001: 288–298.
[67] Vaidyaraman S, Purdy M, Walker T, *et al*. C/SiC material evaluation for aircraft brake applications. In: *High Temperature Ceramic Matrix Composites*. Krenkel W, Naslain R, Schneider H, Eds. Wiley-VCH Verlag GmbH & Co. KGaA, Weinheim, FRG, 2001: 802–808.
[68] Seghi S, Fabio B, Economy J. Carbon/carbon–boron nitride composites with improved wear resistance compared to carbon/carbon. *Carbon* 2004, **42**: 3043–3048.
[69] Seghi S, Lee J, Economy J. High density carbon fibre/boron nitride matrix composites: Fabrication of composites with exceptional wear resistance. *Carbon* 2005, **43**: 2035–2043.
[70] Cai Y, Fan S, Liu H, *et al*. Microstructures and improved wear resistance of 3D needled C/SiC composites with graphite filler. *Compos Sci Technol* 2009, **69**: 2447–2453.
[71] Cai Y, Fan S, Yin X, *et al*. Effects of graphitization degree in three-dimensional needled C/SiC composites on tribological properties. *Int J Appl Ceram Technol* 2011, **8**: 317–328.
[72] Xu X, Fan S, Zhang L, *et al*. Tribological behaviour of three-dimensional needled carbon/silicon carbide and carbon/carbon brake pair. *Tribol Int* 2014, **77**: 7–14.
[73] Cai Y, Xu Y, Li B, *et al*. Low-cost preparation and frictional behaviour of a three-dimensional needled carbon/silicon carbide composite. *J Eur Ceram Soc* 2009, **29**: 497–503.